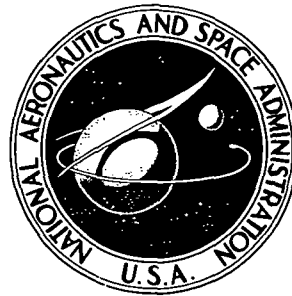


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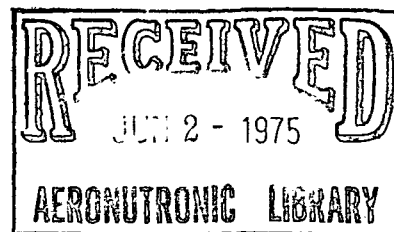


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**RELATION OF STRUCTURE TO
MECHANICAL PROPERTIES OF THIN
THORIA DISPERSION STRENGTHENED
NICKEL-CHROMIUM (TD-NiCr) ALLOY SHEET**

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RELATION OF STRUCTURE TO MECHANICAL PROPERTIES OF THIN THORIA DISPERSION STRENGTHENED NICKEL- CHROMIUM (TD-NiCr) ALLOY SHEET

by John D. Whittenberger

Lewis Research Center

SUMMARY

The relation between structure and mechanical properties of thin TD-NiCr sheet was studied. The results indicate that the elevated temperature tensile, stress-rupture, and creep strength properties depend on grain aspect ratio and sheet thickness. However, they do not depend on grain size when the grain size ranges from 100 to 300 μm . In general, the strength properties increased with increasing grain aspect ratio and sheet thickness. Tensile testing revealed an absence of ductility at elevated temperatures ($T \geq 1144 \text{ K}$). Significant creep damage, as determined by subsequent tensile testing at room temperature, occurred after very small amounts (<0.1 percent) of prior creep deformation at elevated temperatures ($1144 \leq T \leq 1477 \text{ K}$). A threshold stress for creep appears to exist. Creep exposure below the threshold stress at $T \geq 1366 \text{ K}$ resulted in almost full retention of room temperature tensile properties.

INTRODUCTION

Dispersion-strengthened alloys such as TD-NiCr (Ni-20Cr-2ThO₂) are of interest because of their high temperature strength and oxidation resistance. Thin-gage (0.025 to 0.051 cm) TD-NiCr sheet has been extensively evaluated for applications such as the Space Shuttle thermal protection system (ref. 1). In that evaluation, several developmental programs were conducted under NASA sponsorship. They included sheet manufacture (refs. 2 and 3), forming and joining (ref. 4), and property characterization (ref. 5).

In the property characterization study (ref. 5), the mechanical properties of four well-documented heats of thin TD-NiCr sheet were determined. This report correlates the mechanical properties which are reported in reference 5 for thin TD-NiCr sheet

with the structural characteristics of that material. Emphasis is placed on tensile, stress-rupture, and creep behavior as related to microstructure and sheet thickness. In addition, creep damage effects as determined by room temperature tensile testing of as-crept specimens are reported. Evidence for the possible existence of a threshold stress for creep in TD-NiCr sheet is also presented.

W. P. Koster and L. Fritz of Metcut Research Associates, Inc., directed and conducted the majority of the mechanical property tests.

EXPERIMENTAL PROCEDURE

Characteristics of Thin-Gage TD-NiCr Sheet

The TD-NiCr sheet used in the property characterization study (ref. 5) was typical of the sheet produced by Fansteel, Inc., following a standard commercial manufacturing process (refs. 2 and 3). The heat numbers, nominal sheet size, and chemistry of the TD-NiCr sheet are listed in table I. In accordance with standard practice, the 0.025-cm-thick TD-NiCr sheet was finished by belt sanding and was then cold rolled a few percent. The cold-rolling was used to improve the sheet flatness. Previously, it was shown (refs. 2 and 3) that a light cold-roll pass (<5 percent) did not affect the microstructure or mechanical properties of the sheet. The 0.051-cm-thick TD-NiCr sheet was simply finished by belt sanding.

The microstructure of the TD-NiCr sheet was composed of pancake-shaped grains with the short dimension parallel to the sheet thickness. The average grain thickness, grain length, grain size, grain aspect ratio, and ratio of sheet thickness to grain thickness are reported in table II for each heat as a function of orientation (parallel or normal to the sheet rolling direction). While the microstructure of heat 3697 is also composed of pancake-shaped grains, it should be noted that the thickness of the grains is equal to the sheet thickness. Typical microstructures of these TD-NiCr sheets are presented in reference 5.

The microstructural differences noted in table II are due to slight differences in the sheet rolling schedule during sheet manufacture. Manufacture of thin-gage, TD-NiCr sheet required warm pack-rolling (1000 K), which made it difficult to control precisely the reduction per pass of each heat of material. As shown in the sheet manufacturing development studies (refs. 2 and 3), a change in the average reduction per pass from about 5 percent to about 8 percent could produce significant differences in the recrystallized microstructure of TD-NiCr sheet product. Key product specifications, including mechanical properties and formability, could be met even though significant differences in microstructure were obtained.

The (111) pole figures were determined for the TD-NiCr sheet. All four heats possessed an approximate (210) [001] texture with the [001] parallel to the sheet rolling direction. Neither the average thoria particle size nor the average intraparticle spacing were characterized. However, experience with the standard processing schedule (ref. 2) has shown that the average thoria particle size, as determined by X-ray line broadening techniques, ranges from 18 to 20 nm.

Test Methods

The tensile properties of all four heats of material were measured at ambient temperature and at 922, 1144, 1255, 1366, 1477, and 1589 K for specimens taken parallel, at 45°, and normal to the sheet rolling direction. All testing was performed in accordance with ASTM specification E21-69. A sheet-type specimen with a nominal 1.25- by 5.0-cm gage section (1.25- by 6.3-cm reduced section) was used. At all test temperatures a strain rate of 0.005 cm/cm/min was maintained through the 0.2-percent-offset yield; then a crosshead rate of 0.127 cm/min was set and held to failure. Strains through the 0.2-percent-offset yield were measured by a lightweight extensometer which was attached directly to the reduced section of the specimen. Elongations after fracture were measured by normal fitback techniques.

Stress-rupture testing was performed in air on all four heats of TD-NiCr between 1144 and 1589 K. Tests were conducted on tensile-type specimens taken parallel and normal to the sheet rolling direction. All stress-rupture test procedures were in accordance with ASTM specification E139-69.

Constant-load creep tests were performed in air on all four heats of TD-NiCr. Creep extension was measured optically by using interlocking platinum strip extensometers which were tack welded to the reduced section of tensile-type specimens. Tests were performed at 1144, 1255, 1366, and 1477 K on specimens taken parallel and normal to the sheet rolling direction. All creep testing was conducted in accordance with ASTM specification E139-69. Tests were designed to obtain the stresses necessary to produce 0.1 and 0.2 percent total plastic creep in 100 hours of testing. By definition, total plastic creep strain is equal to the plastic creep strain on loading plus the time-dependent strain.

In order to examine the possibility of creep damage, as-crept specimens from heats 3636 and 3712 (0.051 cm thick) were tensile tested at room temperature. The original attempt to obtain such data involved tensile testing of specimens (nominal 1.25- by 6.3-cm reduced section) which had been stripped of the platinum extensometers. Unfortunately, many of these specimens failed at or near the region where the platinum extensometer had been attached. In order to still obtain a measure of the creep damage in the specimens which underwent questionable tensile failure, a nominal 0.5- by 1.9-cm

reduced section was ground into the central portion of the original reduced section, and these specimens were then retested. As a check of this procedure, as-received specimens with the small-gage section were also tested. The tensile properties of these specimens were essentially identical to those measured with the original specimen geometry.

RESULTS AND DISCUSSION

Tensile Properties

The average tensile properties for 0.051- and 0.025-cm-thick sheet as a function of temperature are presented in table III. These data were obtained by averaging test measurements from triplicate tests of each of the three specimen orientations for both heats of each thickness. At any one temperature, the tensile strength properties were found to depend on the heat of material and the specimen orientation. In general, the variation in strength level with respect to each heat of material for each thickness was less than 10 percent of the value reported in table III. The variations in strength level with respect to specimen orientation for any one heat of material were less than 10 percent. Strength levels of specimens taken parallel to the sheet rolling direction were greater than those of specimens taken at 45° to the rolling direction. Also, the latter specimens were, in turn, stronger than specimens taken normal to the rolling direction. The data in table III indicate that the 0.051-cm-thick sheet had slightly better tensile properties than the 0.025-cm-thick sheet. However, both thicknesses exhibited a pronounced lack of ductility at elevated temperatures ($T \geq 1144$ K). This lack of ductility was so severe that tensile fracture occurred before the 0.2-percent-offset yield strength was attained. Previous data for TD-NiCr sheet (refs. 6 and 7) also indicate low elevated temperature tensile ductility, but not as low as that shown in table III.

In general, it is difficult to make direct comparison of mechanical property data for TD-NiCr because of the dependency on prior thermomechanical processing. As previously noted, slight variations in the processing schedule during sheet manufacture can significantly alter the mechanical properties (refs. 2 and 3). However, Wilcox and Clauer (ref. 8) have proposed that various elevated temperature strength properties of dispersion-strengthened alloys depend primarily on the grain aspect ratio. With the assumption that the tensile strengths in table III are reasonable approximations of the 0.2-percent-offset yield strengths, tensile strengths obtained at 1366 K (table III) agreed well with the data shown in figure 11(a) of reference 8 for a grain aspect ratio of about 5 (approximate average for parallel and normal sheet directions of heats 3636, 3637, and 3712).

The low ductility in thin TD-NiCr sheet could possibly be due to the combined effects of a low equicohesive temperature, a low ratio of sheet thickness to grain thickness, and the thinness of the sheet itself. A low equicohesive temperature would be dominant if the thoria particles significantly strengthened the matrix without similarly strengthening the grain boundaries. Following the concept of representing dispersion-strengthened alloys as fibrous composites, as proposed by Frazer and Evans (ref. 9), the strength of the alloy would depend on the grain boundary strength and area. Thus, dispersion-strengthened alloys with small ratios of sheet thickness to grain thickness would have a small amount of grain boundary area and could fail in a brittle manner by intergranular fracture before yielding occurred in the matrix. In addition, there may be an effect on tensile ductility from sheet thickness considerations alone (ref. 10) where the tensile ductility decreases as the sheet thickness decreases.

These concepts tend to be supported by the tensile property data. For example, the tensile ductility of the two heats of 0.051-cm-thick sheet, which possess similar grain aspect ratios, were compared at 922 K. Heat 3636 (sheet thickness/grain thickness ≈ 10) exhibited about 6.5 percent elongation, while heat 3712 (sheet thickness/grain thickness ≈ 5) exhibited about 2 percent elongation. In addition, comparison of the tensile ductility measured at 922 K for heats of TD-NiCr possessing similar grain aspect ratios and ratios of sheet thickness to grain thickness (heats 3637 and 3712) reveals that the thicker material exhibited slightly more ductility (~ 2 percent) than the thinner material (< 1 percent).

Stress-Rupture Properties

Typical stress-rupture data for the thin TD-NiCr sheet are shown in figure 1. Examination of the stress-rupture data for all four heats of thin TD-NiCr sheet showed the following trends:

- (1) At lower test temperatures the 0.051-cm-thick TD-NiCr sheet exhibited greater stress-rupture strength than the 0.025-cm-thick sheet. However, at 1477 K the stress-rupture strengths of the 0.051- and 0.025-cm-thick sheets were essentially equal.
- (2) For any heat of material the stress-rupture strength is greater parallel to the rolling direction than normal to the rolling direction.
- (3) The rupture behavior of both 0.051-cm-thick heats was similar.
- (4) At low test temperatures, heat 3697 tended to exhibit unpredictable rupture behavior where duplicate stress levels produced either instantaneous failure or a life exceeding 300 hours.

The unpredictable behavior of heat 3697 at low temperatures is probably due to single grains traversing the sheet thickness and resulting in rapid intergranular failure in some instances. Since the rupture strengths of both heats of 0.051-cm-thick sheet

are similar. It appears that the stress-rupture properties do not depend on the ratio of sheet thickness to grain thickness or on the grain size, at least for grain sizes between 100 and 300 μm . On the other hand, differences in grain aspect ratio between the two test directions can account for the differences in strength between the parallel and normal directions. Thus, the grain aspect ratio may be the only important microstructural variable governing stress-rupture behavior. Finally, the observation that the 0.025-cm-thick sheet possessed lower rupture properties than the 0.051-cm-thick sheet indicates that the sheet thickness may also affect rupture strength. Similar dependence on specimen size has been observed during stress-rupture testing of nickel-base (refs. 11 and 12) and cobalt-base superalloys (ref. 13).

While the stress-rupture strength does define the maximum strength for long-time, high-temperature conditions, metallographic examination (ref. 14) of stress-rupture-tested TD-NiCr specimens revealed severe microstructural damage in many specimens. Apparently, thin TD-NiCr sheet can undergo massive internal oxidation during testing without failure. In some cases, almost complete conversion to oxide was observed in unfailed test specimens. Wilcox, Clauer, and McCain (ref. 15) have previously shown that such internal oxidation can lead to a considerable increase in rupture life compared to stress-rupture data obtained in vacuum. Thus, stress-rupture data obtained in air for thin TD-NiCr sheet may not be completely reliable for design purposes.

Creep Properties and Threshold Stress

In general, normal creep curves were obtained for all combinations of heat of material, rolling direction, and temperature. (Several typical creep curves and all creep data in the form of measured creep strain as a function of time are given in ref. 5.) In the analysis of the creep data, reference was made to each individual creep curve to obtain the creep strain after 100 hours at the imposed stress level. Approximately six creep curves were available for each heat-direction-temperature combination to provide information on creep strain from about 0.05 percent to a few percent. This information led to 100-hour isochronous stress-strain plots which were then used to determine stresses which would produce either 0.1 or 0.2 percent creep in 100 hours. In figure 2, the stresses necessary to produce 0.1 percent creep in 100 hours are shown as a function of temperature, rolling direction, and sheet thickness. From these data, it can be seen that the creep strength of TD-NiCr sheet was greater when it was tested parallel to the rolling direction rather than normal to the rolling direction and that the 0.051-cm-thick sheet was stronger than the 0.025-cm-thick sheet. Again, because of the equivalence in strength of the two heats of 0.051-cm-thick sheet, the grain aspect ratio seemed to be the most significant microstructural variable. In general, the stresses

required to produce 0.2 percent creep in 100 hours are within 10 percent of the stress values for 0.1 percent creep in 100 hours.

During the development of this creep information, 100-hour isochronous stress-strain diagrams were used. Examination of such plots revealed the existence of an apparent linear relation between applied stress and creep strain after 100 hours of testing. Typical isochronous stress-strain curves are shown in figure 3. Such behavior suggests that thin TD-NiCr sheet behaves as a Bingham solid; that is, the strain rate $\dot{\epsilon}$ is proportional to $\sigma - \sigma_0$, where σ is the applied stress and σ_0 is the threshold stress. Ideally, σ_0 is a stress below which no creep would occur. By linear extrapolation of the isochronous stress-strain curves to zero creep strain, threshold stresses were determined as a function of heat of material, direction, and test temperature. In figure 4 the estimated threshold stresses are shown as a function of temperature for the various heats of material and rolling directions. From figure 4 it is apparent that the threshold stresses depend on sheet thickness, direction, and temperature in the same manner as does the stress to produce 0.1 percent creep. Examination of the threshold stress data for 0.051-cm-thick sheet with respect to the microstructural variables given in table II indicated a possible dependence on grain aspect ratio but not on grain size or the ratio of sheet thickness to grain thickness. Differences in the creep strength between the 0.051- and 0.025-cm-thick sheet indicated, again, a possible dependence of strength on sheet thickness. Finally, the linear relation between threshold stress and temperature was not unexpected. Burton (ref. 16) observed similar behavior in alumina-strengthened copper alloys.

In general, a threshold stress for creep in dispersion-strengthened alloys has been associated with diffusional creep, and evidence for diffusional creep in TD-NiCr has been presented (ref. 14). It has been postulated (refs. 17 to 20) that the threshold stress is due to second-phase particles lying in grain boundaries, where these particles affect the emission and absorption of vacancies. Harris (ref. 19) developed a model for the threshold stress σ_0 based on the punching of prismatic interstitial loops into the matrix alloy, where

$$\sigma_0 \approx \frac{G_m a r}{2\pi\lambda^2} \ln\left(\frac{r}{a}\right) \quad (1)$$

where G_m is the elastic shear modulus of the matrix, a is the interatomic distance, r is the radius of the dislocation loop (assumed to equal that of the dispersoid), and λ is half the intraparticle spacing. Harris also developed an alternative mechanism based on the operation of Frank-Read dislocation sources, where

$$\sigma_0 \approx \frac{G_m b r}{\lambda^2} \quad (2)$$

where b is the Burger vector of the dislocation. A comparison of equations (1) and (2) reveals that they vary only by a small numerical factor. Burton (ref. 20) constructed a model of σ_0 based on the relaxation of stresses by the nucleation and growth of defect loops in the particle/matrix interface in which

$$\sigma_0 = \frac{\pi U^2 V}{55 b k T} \quad (3)$$

for large particles (particle radius > 10 nm); where U is the line energy, V is the volume fraction of particles, k is Boltzmann's constant, and T is the absolute temperature. Burton suggests that $U \propto (G_m + G_p)$, where G_p is the shear modulus of the particle. Therefore,

$$\sigma_0 \propto \frac{(G_m + G_p)^2}{T} \quad (4)$$

Examination of equations (1) to (4) reveals that the only temperature-sensitive terms are G_m and $(G_m + G_p)^2/T$. Thus, the theoretical temperature dependence of the threshold stress can be calculated from data on the shear modulus as a function of temperature. With the assumptions that the shear modulus of Ni-20Cr is approximately equal to that of TD-NiCr (ref. 21) and that Poisson's ratio is not a function of temperature, G_m can be estimated from the elastic modulus and Poisson's ratio data for TD-NiCr given in reference 5. As the elastic modulus data for TD-NiCr decrease linearly between 922 and 1366 K (the highest temperature evaluated), the shear modulus decreases linearly with temperature. The shear modulus of thoria (ref. 22) also decreases linearly with temperature, and all three threshold stress models agree with the observed temperature dependence of the threshold stress. The value of $\sigma_0(1366 \text{ K})/\sigma_0(1144 \text{ K})$ as determined experimentally from the creep data for heats 3636, 3712, and 3637 ranged from 0.45 to 0.55. Theoretical calculations based on equation (1) or (2) gave a value of 0.75, and similar calculations using equation (4) gave a value of 0.50. From these data, it appears that the temperature dependence of Burton's model more closely fits the observed behavior.

Finally, it is possible to calculate theoretical values of the threshold stress from equations (1) and (2); a similar calculation based on equation (3) is not possible because of the unknown relation between U and $G_m + G_p$. For $2r = 14.5 \text{ nm}$, $2\lambda = 1.32 \text{ } \mu\text{m}$

(data from ref. 15), $a = 0.35 \text{ nm}$, $b = 0.25 \text{ nm}$, $G_m(1144 \text{ K}) \approx 40 \text{ GN/m}^2$, and $G_m(1366 \text{ K}) \approx 30 \text{ GN/m}^2$; equations (1) and (2) yield threshold stresses of about 15 MN/m^2 at 1144 K and about 10 MN/m^2 at 1366 K . Comparison of the calculated values with those shown in figure 4 indicates that the calculation yields about the right order of magnitude.

Creep Damage

One major area of concern in the use of thin TD-NiCr sheet is the apparent loss of room temperature tensile properties after high temperature creep (ref. 11). In order to confirm the possibility of creep damage, as-crept specimens from heats 3636 and 3712 were tensile tested at room temperature. Typical results for the residual tensile properties of as-crept specimens (heat 3712 tested normal to the rolling direction) are shown in figure 5. This figure illustrates the drastic loss in room temperature tensile ductility (elongation) and the accompanying loss in tensile strength which occur after small creep deformations at temperatures ranging from 1144 to 1477 K .

In general, the degradation of tensile properties appears to depend on the amount of prior creep strain and not on the temperature of prior creep testing. Both heats and both testing directions exhibited similar behavior. The test data for all residual property tests are given in the appendix. The retest procedure tended to yield higher tensile properties than the original residual tensile tests when data from as-crept specimens with similar creep histories were compared. Such behavior was not unexpected as creep deformation in TD-NiCr appears to be inhomogeneous (ref. 11). Also, creep damage effects may be more prevalent at the edges of the original specimens.

It has been proposed (ref. 11) that creep damage in TD-NiCr is due to diffusional creep (Nabarro-Herring and/or Coble model) which forms thoria-free regions around grain boundaries generally lying perpendicular to the applied tensile stress. Such thoria-free regions then can act as sites for voids and intergranular oxidation and produce further creep degradation. An alternative diffusional creep model has been proposed by Risihi Raj (personal communication), who contends that the applied tensile stress first produces voids on the grain boundaries. Once formed, the voids act as vacancy sinks and the surrounding grain boundaries act as vacancy sources. Thus, thoria-free regions are produced at grain boundaries near the voids.

Since the creep-damaged regions are much weaker than the surrounding materials, subsequent tensile deformation would preferentially occur in these regions. In order to determine if such creep damage did occur, the tensile fracture surfaces of heat 3712 (normal) creep tested at 1366 K were examined with aid of the scanning electron microscope. The prior creep histories and residual tensile properties of the specimens are given in table IV.

Examination of the specimens from the residual property tests revealed three distinct structures:

- (1) Type I - the as-received fracture surface, which is composed of relatively flat regions containing occasional small pits and rougher regions containing close-set linear arrays of small pits (probably ductile fracture pits initiating on Cr_2O_3 stringers)
- (2) Type II - the oxidized fracture surface
- (3) Type III - the lacy-network surface which is formed by the outline of very large fracture dimples

Typical examples of these fracture surfaces are shown in figure 6, and the types of fracture surfaces observed for each specimen are listed in table IV.

A type I fracture surface was required for full retention of the original room temperature tensile properties; a combination of types I and III generally meant an intermediate retention; and a combination of all types or type II alone meant essentially the complete loss of room temperature ductility. The lacy-network (type III) fracture surface is believed to be the result of tensile failure by microvoid coalescence and tearing in a creep-damaged region. It is quite possible that the very large size of the fracture dimples is due to the presence of creep voids in the thoria-free regions.

During the course of this program, specimens from heat 3712 were creep tested at stress levels equal to and slightly below the threshold stresses shown in figure 4. In general, these specimens exhibited small (<0.1 percent) creep deformations after testing. In view of the experimental problems when testing at elevated temperatures and low strains, it was difficult to assess whether or not the measured creep deformations were real. In order to determine if creep damage had occurred in these test specimens, the residual room temperature tensile properties were also measured.

The creep histories and residual properties of creep specimens from heat 3712 tested at and slightly below the estimated threshold stresses are presented in table V. In general, it appears that almost complete retention of room temperature tensile properties after creep exposures at 1477 and 1366 K can be expected for stresses less than or equal to σ_0 . However, creep exposures at or slightly below estimated threshold stresses at 1144 and 1255 K apparently produced creep damage and hence the tendency for low residual properties. The difficulty of accurately assessing the threshold stress is probably due to the inhomogeneity of creep in TD-NiCr. If a thoria-free region were produced anywhere in the reduced section during creep exposure, subsequent tensile deformation would be expected to occur preferentially within that region.

CONCLUDING REMARKS

Measurements of the mechanical properties of two heats each of 0.051-cm-thick and 0.025-cm-thick TD-NiCr sheet have shown that elevated temperature tensile, stress-

rupture, and creep strength properties depend primarily on grain aspect ratio and sheet thickness. In general, it was found that the strength properties increased with increasing grain aspect ratio and increasing sheet thickness. Grain size, at least for TD-NiCr sheet with an average grain size between 100 and 300 μm , did not appear to be an important strength factor at elevated temperatures.

Additionally, this study of thin TD-NiCr sheet revealed (1) that it apparently has nil tensile ductility at elevated temperatures; (2) that even very small amounts of prior creep deformation (<0.1 percent) at elevated temperatures ($T \geq 1144$ K) can produce severe creep damage, which is responsible for the low residual tensile properties; and (3) that there appears to be a threshold stress for creep.

The observation of nil elevated temperature tensile ductility and the loss of room temperature tensile properties after creep exposure seem to place severe limitation on the use of thin TD-NiCr sheet. However, it appears that the problem of low residual properties could be avoided in structural applications by using the material at stress levels that are below the threshold stress. It should be noted that the threshold stresses for TD-NiCr sheet are relatively high: for example, at 1366 K the threshold stress for 0.051-cm-thick sheet tested normal to the rolling direction is about 35 MN/m^2 , which is approximately 70 percent of the 100-hour rupture stress. Thus, thin TD-NiCr sheet possesses a useful strength capacity at high temperatures. Finally, because of the apparent relation of high temperature - low strain rate deformation in dispersion-strengthened alloys with diffusional creep mechanisms (ref. 11), there is no reason to suspect that creep damage effects and threshold stresses are unique to thin TD-NiCr sheet.

While this report discusses only the tensile, stress-rupture, and creep properties of thin TD-NiCr sheet, other mechanical properties of this material were measured as part of this property study. They include modulus of elasticity, Poisson's ratio, bearing strength, compression, shear strength, sharp notch, and fatigue. In addition, physical properties including linear thermal expansion, thermal conductivity, specific heat, total hemispherical emittance, thermal diffusivity, and electrical resistivity were also determined for TD-NiCr sheet. All mechanical and physical property data for thin TD-NiCr sheet are reported in reference 5.

CONCLUSIONS

Based on a study of the mechanical properties and structure of thin TD-NiCr sheet (0.051 and 0.025 cm thick), the following conclusions were drawn:

1. Elevated temperature tensile, stress-rupture, and creep strength properties depend on grain aspect ratio and sheet thickness but apparently not on grain size when the grain size ranges from 100 to 300 μm .

2. Thin TD-NiCr sheet apparently has nil tensile ductility at elevated temperatures ($T \geq 1144$ K).

3. Significant creep damage, as determined by residual property testing at room temperature, occurs after very small amounts (<0.1 percent) of prior creep deformation ($1144 \leq T \leq 1477$ K).

4. A threshold stress for creep appears to exist. For 0.051-cm-thick TD-NiCr sheet, the threshold stress is about 40 MN/m^2 at 1366 K. Creep exposures below the threshold stress result in almost full retention of the residual tensile properties.

Lewis Research Center,
National Aeronautics and Space Administration,
Cleveland, Ohio, March 7, 1975,
506-16.

APPENDIX - RESIDUAL PROPERTY TESTS

Following creep testing, the specimens from heats 3712 and 3636 were tensile tested at room temperature to determine the extent of creep damage. As-crept specimens were stripped of the platinum creep extensometer and then tensile tested at room temperature following the test procedure previously outlined in the main text. Tensile testing was done at Metcut Research Associates, Inc., under NASA purchase order C-38303H-C. Unfortunately, many of these specimens failed at or near the regions where the extensometer had been originally attached. In order to at least obtain an estimate of the creep damage, a small, nominally 0.5 cm by 1.9 cm, reduced section was ground into the central portion of the specimens which had undergone questionable failure. These were then retested at 0.127-cm/min crosshead speed to failure. This retesting procedure was conducted at the Lewis Research Center; only ultimate tensile strength and elongation at fracture were recorded. The yield strength at 0.02 and 0.2 percent offset could not be accurately determined because of the original attempt to measure the residual properties. The creep histories and residual properties of specimens from heats 3636 and 3712 are presented in tables VI and VII. The room temperature tensile properties of as-received TD-NiCr as determined with the small reduced section are given in table VIII.

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TABLE I. - NOMINAL SHEET SIZE AND
CHEMISTRY OF TD-NiCr

Heat	Cr	C	S	ThO ₂	Nominal sheet size, m	Nominal thickness, cm
	Composition, wt% (balance Ni)					
3636	20.16	0.0214	0.0021	2.17	0.5 by 0.9	0.051
3637	19.8	.0194	.0019	2.14	0.45 by 0.9	.025
3697	20.07	.0248	.0055	2.10	0.45 by 0.9	.025
3712	19.77	.025	.003	2.10	0.45 by 1.05	.051

TABLE II. - AVERAGE GRAIN-SIZE PARAMETERS OF
THIN TD-NiCr SHEET

Heat	Orientation, in reference to sheet roll- ing direction	Grain thickness, D, μm	Grain length, L, μm	Grain size, \sqrt{LD} , μm	Grain aspect ratio, L/D	Ratio of nominal sheet thickness to average grain thickness
3636	Parallel	50	290	120	6	10.2
	Normal	50	230	110	4.5	10.2
3637	Parallel	50	390	140	7.5	5.1
	Normal	55	290	130	5	4.6
3697	Parallel	250	---	---	---	1.0
	Normal	250	---	---	---	1.0
3712	Parallel	110	640	265	6	4.6
	Normal	90	360	180	4	5.6

TABLE III. - AVERAGE TENSILE PROPERTIES OF THIN TD-NiCr
SHEET AS FUNCTION OF SHEET THICKNESS AND TEMPERATURE

Temperature, K	Sheet thickness, cm							
	0.051	0.025	0.051	0.025	0.051	0.025	0.051	0.025
	Yield strength at 0.02 per- cent offset, MN/m ²		Yield strength at 0.2 percent offset, MN/m ²		Ultimate ten- sile strength, MN/m ²		Elongation, percent	
297	478	516	570	612	845	798	17.3	11.2
922	271	309	317	(a)	338	360	4.0	.7
1144	^b 160	(a)	^b 179	(a)	182	162	1.5	.6
1255	^b 136	(a)	(a)	(a)	146	118	.9	.5
1366	^b 104	(a)	(a)	(a)	116	88	.8	.5
1477	(a)	(a)	(a)	(a)	92	72	.9	.5
1589	(a)	(a)	(a)	(a)	75	56	.7	.4

^aMore than two test failures prior to reaching offset yield.

^bOne or two of 18 test specimens fractured prior to reaching offset yield.

TABLE IV. - RESIDUAL PROPERTIES OF TD-NiCr SHEET
AFTER HIGH TEMPERATURE CREEP

[Heat 3712 tested normal to rolling direction at 1366 K.]

Specimen	Creep history			Residual tensile properties					Residual fracture surface type ^a
	Stress, MN/m ²	Creep, percent	Time, hr	Original test		Retest		Overall tensile elongation, percent	
				Ultimate tensile strength, MN/m ²	Tensile elongation, percent	Ultimate tensile strength, MN/m ²	Tensile elongation, percent		
TB-7-4 ^b	----	----	-----	---	---	784	25	25	I
CJ-7-45	27.6	0.01	103.3	725	7.8	842	14	22	I
CJ-1-35	34.5	.02	111.0	445	.8	---	--	1	I, II, III
CJ-5-52	27.6	.05	117.1	701	6.1	833	16	22	I
CJ-5-22	37.9	.08	115.8	347	1.2	713	7	8	I, III
CJ-7-44	34.5	.15	113.0	597	1.8	777	13	15	I, III
CJ-7-31	41.4	.22	99.4	274	.9	---	--	1	I, II, III
CJ-1-16	41.4	.26	102.1	274	.7	368	1	1.5	I, II, III
CJ-8-8	43.4	.33	121.6	273	2.2	---	--	2	I, II, III
CJ-1-51	39.4	.45	119.8	231	1.2	292	0	1	I, II, III
CJ-5-9 ^c	43.3	4.1	121.6	---	---	---	--		II

^aFracture surface types are described in text (p. 10).

^bAs received.

^cFailed during unloading.

TABLE V. - CREEP HISTORIES AND RESIDUAL ROOM TEMPERATURE TENSILE PROPERTIES
OF TD-NiCr SHEET CREEP TESTED AT AND BELOW THRESHOLD STRESS

[Heat 3712.]

Specimen	Creep history				Specimen	Creep history				Residual tensile properties		Residual tensile properties											
	Stress, MN/m ²	Time, hr	Strain, percent	Temperature, K		Stress, MN/m ²	Time, hr	Strain, percent	Temperature, K	Ultimate tensile strength, MN/m ²	Elongation, percent	Ultimate tensile strength, MN/m ²	Elongation, percent										
Parallel to rolling direction												Normal to rolling direction											
CJ-8-27	110.3	112.0	0.1	1144	677	3	CJ-3-53	68.9	112.2	0.1	1144	725	9										
CJ-6-44	110.9	123.2	.08	↓	829	11	CJ-7-47	68.9	99.8	.07	↓	588	1.5										
CJ-3-45	103.4	101	.05	↓	683	3.3	CJ-6-51	62.1	100.0	.04	↓	708	8										
CJ-8-29	103.4	100	.08	↓	731	8	CJ-3-52	62.1	113.1	.04	↓	733	11										
CJ-7-41	79.3	100.6	.08	1255	590	1.5	CJ-7-43	51.7	100.0	.06	1255	610	2										
CJ-5-44	79.3	100	.04	↓	790	8.5	CJ-4-50	51.7	123.4	.12	↓	604	2										
CJ-8-26	72.4	101	.02	↓	603	1.5	CJ-7-46	44.8	99.8	.07	↓	777	2										
CJ-2-45	72.4	113.2	.06	↓	855	20	CJ-8-10	44.8	100.2	.03	↓	677	4.5										
CJ-7-37	48.3	101.3	.01	1365	816	11	CJ-1-35	34.5	111.0	.02	1366	445	1										
CJ-7-40	48.3	112.1	.01	↓	871	19	CJ-7-44	34.5	113.0	.15	↓	777	15										
CJ-7-36	41.4	99.9	.15	↓	903	15	CJ-5-52	27.6	117.1	.05	↓	833	22										
CJ-8-28	41.4	100.4	.1	↓	884	17.5	CJ-7-45	27.6	103.3	.01	↓	842	22										
CJ-2-8	31.0	94.5	.02	1477	856	21	CJ-4-45	24.1	100.0	.1	1477	752	17										
CJ-4-42	31.0	117.1	.1	↓	808	10.5	CJ-5-51	24.1	101.2	.15	↓	752	14										
CJ-6-17	24.1	118.3	.05	↓	875	17.5	CJ-4-49	17.2	111.2	.01	↓	809	23										
CJ-7-35	24.1	106.3	.03	↓	845	16.5	CJ-1-53	17.2	113.1	.01	↓	795	25										
As received	-----	-----	-----	-----	872	16.5	As received	-----	-----	-----	-----	801	21.5										

TABLE VI. - CREEP HISTORIES AND RESIDUAL TENSILE PROPERTIES
OF TD-NiCr SPECIMENS FROM HEAT 3636

(a) Prior creep at 1144 K

Specimen	Creep history			Residual tensile properties			
	Stress, MN/m ²	Strain, percent	Time, hr	Ultimate tensile strength, MN/m ²	Tensile elongation, percent		
					Original	Retest	Total
Parallel to rolling direction							
CD-1-8	110.3	0.09	141.5	833	9.1	--	9
CD-1-46	117.2	.13	109.3	873	4.3	9	13
CD-3-45	120.7	.22	100.0	742	3.7	--	4
CD-2-20	124.1	.40	123.0	632	.3	4	4
CD-3-8	127.6	.30	123.3	558	.6	--	.5
CD-4-40	131.0	.75	116.1	455	.4	1	1.5
CRD-1-37	124.1	^a .5	925.0	394	.8	1	2
Normal to rolling direction							
CD-1-40	72.4	0.16	125.6	635	1.9	--	2
CD-2-17	75.8	.16	118.1	582	1.2	--	1
CD-2-37	79.3	.12	116.3	572	1.1	--	1
CD-2-62	80.7	.24	114.7	485	1.1	--	1
CD-3-39	82.7	.51	100.0	468	.3	--	.5
CD-1-12	82.7	.98	115.0	453	1.1	1	2
CRD-2-63	100.0	^a 2.2	571.5	229	1.0	--	1

(b) Prior creep at 1255 K

Parallel to rolling direction							
CD-1-16	48.3	0.06	133.8	936	12	2	14
CD-3-51	82.7	.10	116.5	850	1.0	11	12
CD-5-46	86.2	.12	110.6	764	.6	7	7.5
CD-5-20	89.6	.16	101.9	785	1.0	6	7
CD-4-42	89.6	.69	112.7	365	.9	1	2
CD-3-16	96.5	1.73	122.8	202	.9	--	1
Normal to rolling direction							
CD-1-24	41.4	0.06	121.6	899	8.3	13	21
CD-2-19	51.7	.08	101.4	581	1.3	--	1.5
CD-2-39	55.2	.15	100.0	664	.7	3	4
CD-3-47	56.5	1.49	117.2	326	.1	--	0
CD-3-12	58.6	.91	118.3	439	1.2	1	2
CD-1-49	62.1	2.85	308.6	230	----	1	1

(c) Prior creep at 1366 K

Parallel to rolling direction							
CD-4-4	55.2	0.08	117.7	898	----	20	20
CD-2-42	48.3	.01	101.2	911	----	17	17
CD-1-36	31.0	.02	118.2	940	----	19	19
CD-4-48	62.1	.05	110.1	914	----	13	13
CD-5-4	65.5	.31	125.9	265	----	0	0
Normal to rolling direction							
CD-2-57	33.1	0.08	101.2	832	----	15	15
CD-4-44	36.5	.22	100.0	687	----	1	1
CD-3-57	37.9	1.27	101.5	150	----	1	1
CD-2-29	35.9	.16	120.0	792	-----	11	11

(d) Prior creep at 1477 K

Parallel to rolling direction							
CD-2-44	31.0	0.02	98.5	910	5.3	13	18
CD-1-44	37.9	.09	110.7	826	8.2	--	8
CD-4-20	39.3	.16	127.9	783	.4	15	16
CD-5-45	42.7	.04	125.9	884	5.9	9	15
Normal to rolling direction							
CD-2-59	20.7	0.04	99.9	704	3.1	--	3
CD-1-30	22.1	.10	110.8	860	2.5	15	17
CD-3-29	27.1	.21	110.6	801	1.4	7	8
CD-1-63	23.4	.32	128.0	715	.5	8	8

^aEstimated.

TABLE VII. - CREEP HISTORIES AND RESIDUAL TENSILE PROPERTIES OF TD-NiCr SPECIMENS FROM HEAT 3712

(a) Prior creep at 1144 K

Specimen	Creep history			Residual tensile properties			
	Stress, MN/m ²	Strain, percent	Time, hr	Ultimate tensile strength, MN/m ²	Tensile elongation, percent		
					Original	Retest	Total
Parallel to rolling direction							
CJ-6-25	113.8	0.07	101.2	755	2.4	4	6
CJ-4-8	124.1	.25	101.4	575	1.0	--	1
CJ-1-27	127.6	.41	113.1	493	.2	2	2
CJ-2-18	131.0	.44	113.7	308	1.0	--	1
CJ-7-32	134.4	1.05	137.0	295	.2	2	2
CJ-8-27	110.3	.1	112.0	677	2.8	--	2.8
CJ-6-44	110.9	.08	123.2	829	1.9	9	11
CJ-3-45	103.4	.05	101.1	683	3.5	--	3.5
CJ-8-29	103.4	.08	100.0	731	2.8	5	8
CRJ-1-30	134.4	^a 1.0	448.0	458	.7	--	1
CRJ-5-37	128.9	^a 1.0	449.7	358	.2	--	0
CRJ-3-17	127.6	^a .8	498.6	477	.6	--	.5
CJ-5-19	141.3	1.2	116.0	447		3	3
Normal to rolling direction							
CJ-4-22	75.8	0.12	100.1	538	0.9	--	1
CJ-5-29	75.8	.16	119.8	493	.5	2	2
CJ-3-4	82.7	.26	136.8	483	.9	--	1
CJ-7-21	82.7	.25	117.5	475	.8	--	1
CJ-2-3	89.6	.61	122.5	615	1.2	2	3
CJ-1-9	103.4	.84	101.3	234	1.6	--	1.5
CJ-3-53	68.9	.1	112.2	725	1.0	8	9
CJ-7-47	68.9	.07	99.8	588	1.5	--	1.5
CJ-6-51	62.1	.04	100.0	708	8.2	--	8
CJ-3-52	62.1	.04	113.1	733	----	11	11

(b) Prior creep at 1255 K

Parallel to rolling direction							
CJ-1-31	82.7	0.08	113.3	494	0.2	--	0
CJ-4-14	89.6	.14	116.6	674	.6	6	6.5
CJ-7-8	93.1	.24	101.7	523	.6	--	.5
CJ-7-15	93.1	.30	117.5	612	1.2	4	5
CJ-8-6	100.0	.49	99.7	438	.6	--	.5
CJ-5-31	103.4	.98	101.1	440	1.2	--	1
CJ-7-41	79.3	.08	100.6	590	1.5	--	1.5
CJ-5-44	79.3	.04	100.0	790	4.6	4	8.5
CJ-8-26	72.4	.02	101.1	603	1.6	--	1.5
CRJ-2-40	101.4	^a 1.5	447.4	37	.2	--	0
CRJ-1-41	100.1	^a .8	473.4	384	.5	1	1
CRJ-1-26	96.5	^a 1.2	503.1	381	.4	1	1
CJ-2-45	72.4	.06	113.2	854	----	20	20
Normal to rolling direction							
CJ-2-20	55.2	0.1	101.9	545	1.5	--	1.5
CJ-6-13	55.2	.15	101.2	489	.9	--	1
CJ-7-27	58.6	.20	116.4	306	.7	2	2.5
CJ-3-28	58.6	.29	138.7	467	1.1	--	1
CJ-4-30	65.5	.54	99.8	443	.7	3	3.5
CJ-1-12	68.9	.79	101.2	309	.5	--	.5
CJ-7-43	51.7	.06	100.0	610	1.8	--	2
CJ-4-50	51.7	.12	123.4	604	1.0	1	2
CJ-7-46	44.8	.07	99.8	777	1.1	1	2
CJ-8-10	44.8	.03	100.2	677	4.4	--	4.5
CRJ-6-12	75.8	^a 2.3	400.7	246	.9	--	1
CRJ-4-47	70.3	^a 2.4	495.3	258	1.1	--	1
CRJ-6-48	68.9	^a 1.5	405.5	309	.8	--	1

^aEstimated.

(c) Prior creep at 1366 K

Specimen	Creep history			Residual tensile properties			
	Stress, MN/m ²	Strain, percent	Time, hr	Ultimate tensile strength, MN/m ²	Tensile elongation, percent		
					Original	Retest	Total
Parallel to rolling direction							
CJ-1-38	44.8	0.02	113.4	811	0.3	13	13
CJ-6-7	62.1	.06	101.4	817	.9	16	17
CJ-8-13	68.9	.29	119.2	372	.2	2	2
CJ-7-37	48.3	.01	101.3	816	2.4	9	11.5
CJ-7-36	41.4	.15	99.9	903	12.9	2	15
CJ-8-28	41.4	.10	100.4	884	9.6	8	18
CJ-7-40	48.3	.01	112.1	871	----	19	19
Normal to rolling direction							
CJ-7-45	27.6	0.01	103.3	842	7.8	14	22
CJ-1-31	34.5	.02	111.0	445	.8	--	1
CJ-5-52	27.6	.05	117.1	833	6.1	16	22
CJ-2-22	37.9	.08	115.8	713	1.2	7	8
CJ-7-44	34.5	.15	113.0	777	1.8	13	15
CJ-3-31	41.4	.22	99.4	274	.9	--	1
CJ-1-16	41.4	.26	102.1	368	.7	1	1.5
CJ-8-8	43.4	.33	121.6	273	2.2	--	2
CJ-1-51	39.4	.45	119.8	292	1.2	0	1

(d) Prior creep at 1477 K

Parallel to rolling direction							
CJ-4-36	31.0	0.03	100.3	850	7.5	13	20
CJ-1-43	34.5	.05	102.2	842	4.8	13	18
CJ-5-40	37.9	.07	123.0	868	.9	19	20
CJ-8-7	41.4	.03	115.5	853	5.2	20	25
CJ-2-8	31.0	.02	94.5	856	10.0	11	21
CJ-4-42	31.0	.1	117.1	808	10.4	--	10.5
CJ-6-17	24.1	.05	118.3	875	9.5	8	17.5
CJ-7-35	24.1	.03	106.3	845	7.4	9	16.5
CJ-6-24	42.7	.46	99.7	502	----	1	1
Normal to rolling direction							
CJ-4-45	24.1	0.1	100.0	752	0.8	16	17
CJ-5-12	25.5	.04	100.3	79.6	.4	--	.5
CJ-3-50	25.5	.16	129.3	690	.2	2	2
CJ-8-15	27.6	.33	102.1	568	.9	2	3
CJ-5-51	24.1	.15	101.2	752	1.0	13	14
CJ-4-49	17.2	.01	111.2	809	7.8	15	23
CJ-1-53	17.2	.01	113.1	795	4.0	21	25

TABLE VIII. - ROOM TEMPERATURE TENSILE
 PROPERTIES OF AS-RECEIVED TD-NiCr
 SHEET AS DETERMINED WITH SMALL
 NOMINAL 0.5-cm BY 1.9-cm
 REDUCED SECTION

Specimen	Ultimate tensile strength, MN/m ²	Tensile elongation, percent
Heat 3636 tested normal to rolling direction		
CD-5-21	835	23
CD-4-7	833	23
CRD-3-56	831	23
CRD-1-50	867	21
Average	842	23
Average (ref. 5)	846	21
Heat 3712 tested normal to rolling direction		
CJ-8-16	775	25
CRJ-2-51	798	24
T-J-7-4	784	25
CJ-6-4	795	27
Average	788	25
Average (ref. 5)	801	22

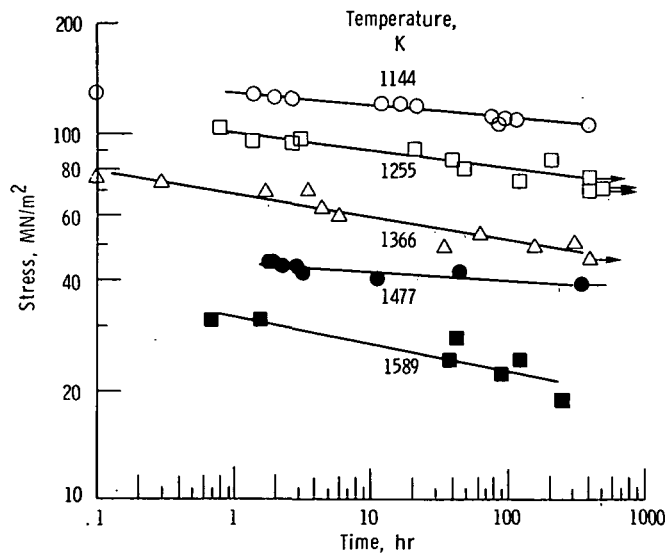


Figure 1. - Stress-rupture strength as a function of temperature. Heat 3712 tested normal to rolling direction; sheet thickness, 0.051 cm.

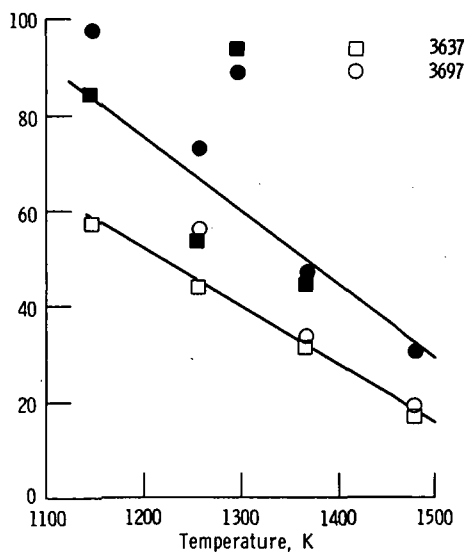
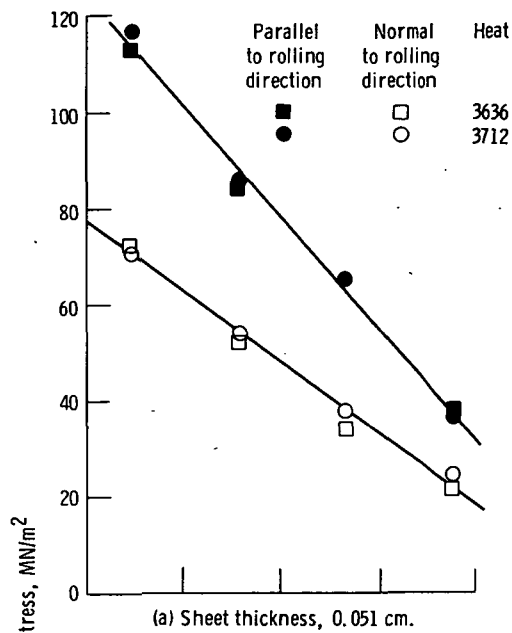


Figure 2. - Stress to produce 0.1 percent creep in 100 hours in thin TD-NiCr sheet as function of temperature and direction.

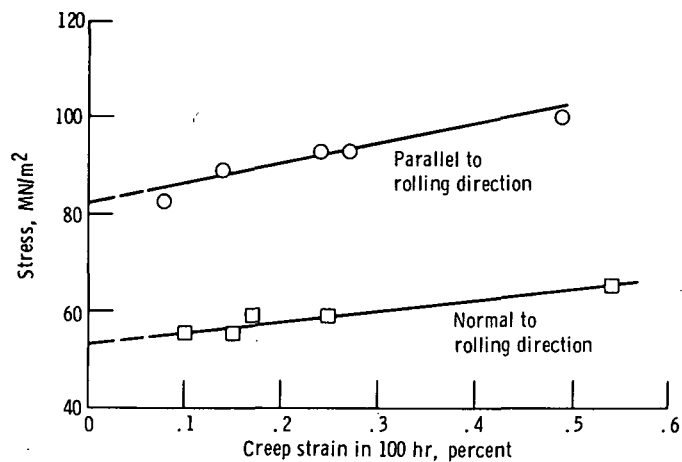


Figure 3. - Isochronous stress-strain behavior for thin TD-NiCr sheet. Heat 3712 creep tested at 1255 K.

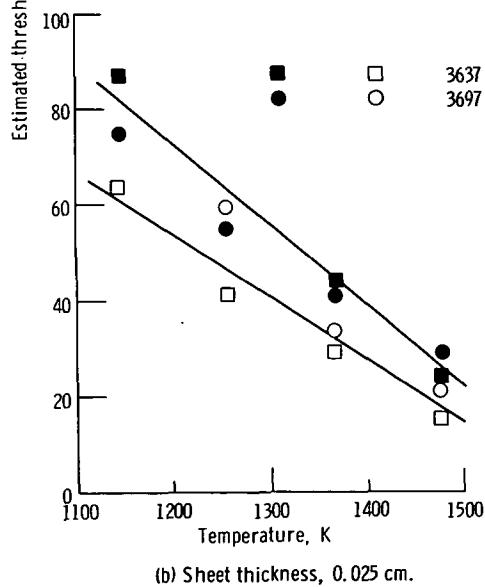
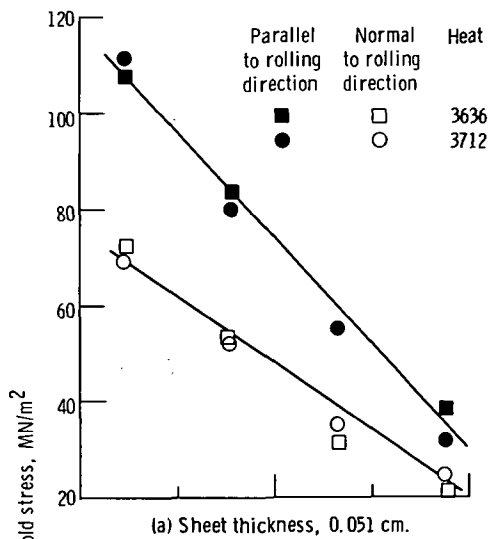


Figure 4. - Estimated threshold stress of thin TD-NiCr sheet as function of temperature and direction.

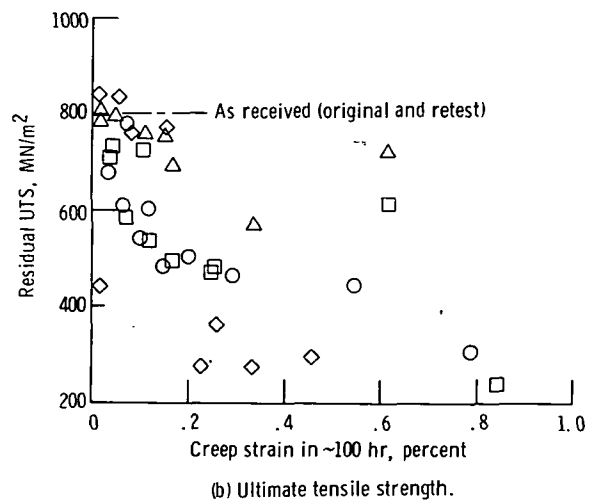
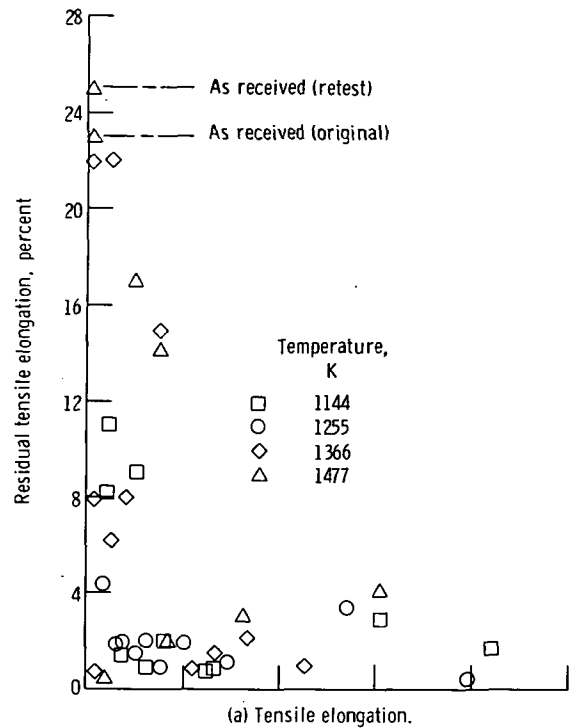
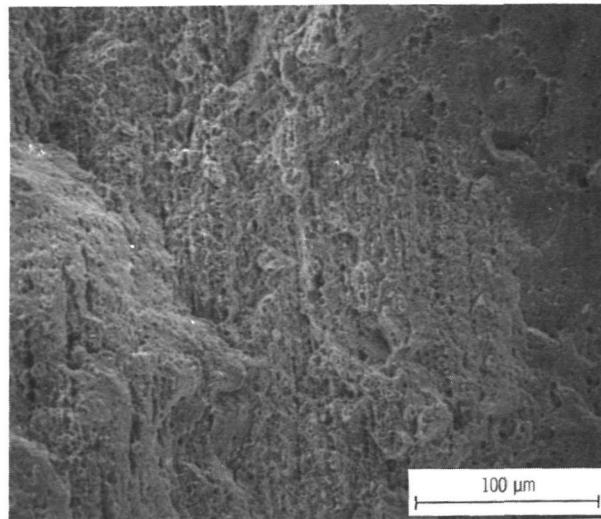
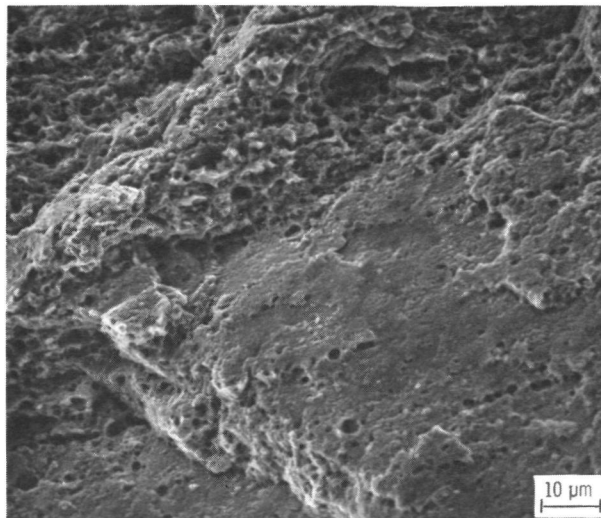


Figure 5. - Residual room temperature tensile properties of thin TD-NiCr sheet as a function of prior creep strain. Heat 3712 tested normal to rolling direction after creep testing at various temperatures for approximately 100 hours.

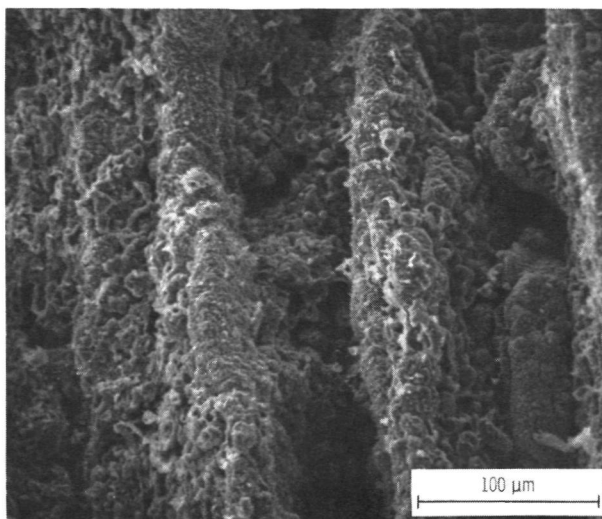


(a) Type I fracture surface; specimen CJ-7-45.

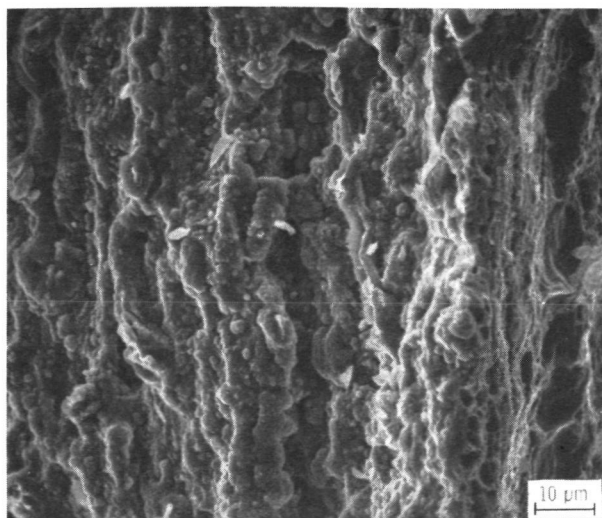


(b) Type I fracture surface; specimen TJ-7-4.

Figure 6. - Typical fracture surfaces observed on thin TD-NiCr sheet specimens used for residual property tests. (See table IV for description of creep testing and residual property data.)

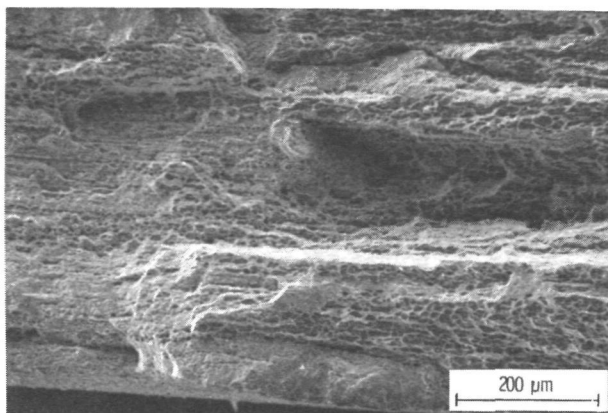


(c) Type II fracture surface; specimen CJ-5-9.

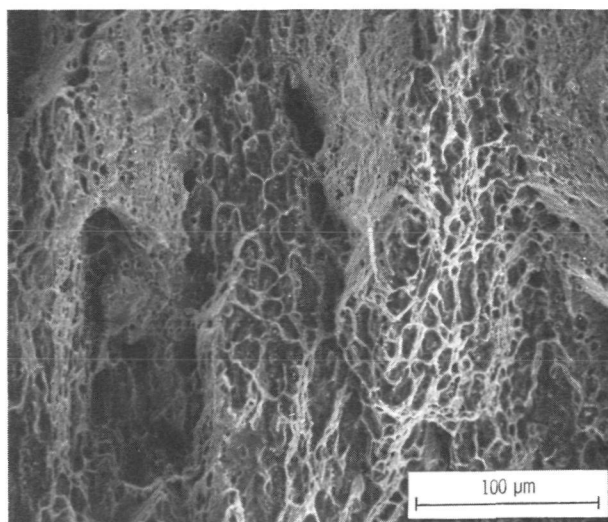


(d) Type II fracture surface; specimen CJ-1-35.

Figure 6. Continued.



(e) Type III fracture surface; specimen CJ-2-22.



(f) Type III fracture surface; specimen CJ-1-35.

Figure 6. - Concluded.



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